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Predicting the upper-bound of interlaminar impact damage in structural composites through a combined nanoindentation and computational mechanics technique

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ABSTRACT

Low-energy/speed impact damage of composite laminates is still challenging to simulate due to difficulties in measuring some key material properties. The present study develops an integrated numerical and experimental method for predicting interlaminar impact damage. A nanoindentation technique for measuring the stiffness properties of composites at a small length scale (nanometers) is leveraged to determine damage to composite laminates due to fast (microsecond) projectile impact. Specifically, nanoindentation is employed to measure the contact stiffness of an indenter and aerospace carbon/epoxy IM7/977-3 laminates with four different stacking sequences. Then, through a technique that combines nanoindentation and computational mechanics, an equivalent impact force approach is proposed to predict the upper-bound of interlaminar impact damage at impact energy levels of 5 and 10 Joules. Drop-weight impact experiments are conducted to validate the prediction results. In practical applications, estimating the upper-bound of damage is important for conservative and efficient damage tolerance designs, especially for thick composite laminates.

1. Introduction

Composite materials have been used increasingly as structural components in aerospace and other industries. Despite their favorable and highly-controllable directional properties, their resistance against out-of-plane impact is still a major concern [2,11]. Especially, low-energy impact events such as those arising from tools dropped during routine maintenance or hailstorms can cause invisible or barely visible impact damage (BVID), which can lead to in-service failure [67,13,33,14,62,24,72,53,32]. Hence, for more than three decades, researchers have developed experimental/numerical techniques for investigating the impact of structural composites [5,52].

In impact experiments using high-speed photography, Lambros and Rosakis [34] studied dynamic delamination of thick composite laminates. Xu and Rosakis [60] focused on the dynamic failure of model sandwiches. Xu and Rosakis [61] also conducted low-speed impact tests on a model composite specimen using ultra-high-speed photography (an impact video will be provided as an online supplemental material). After a spherical steel projectile impacted a model matrix layer (Homalite, which is a transparent polymer) of the specimen, stress wave propagation occurred first, followed by elastic indentation. Clearly, indentation mechanics played an important role in composite projectile impact, which is also referred to as "dynamic indentation" [2]. The third event was a complicated impact damage process, which is the subject considered here for predicting the impact damage.

In terms of impact simulations, Geubelle and Baylor [23] introduced a numerical method for the delamination process in thin composite plates using a two-dimensional cohesive element scheme. Johnson et al. presented a material failure model capable of predicting the intra- and inter-laminar damage. Olsson et al. [40] investigated the growth of delamination at the most critical interface of a laminate. In recent years, numerous impact analysis frameworks integrating damage modeling and experiments have been introduced [35,36,9,69,3,68,50,19,28,15]. Nominally, predicting/simulating impact damage is a challenging task.
as it requires validation at highly resolved spatiotemporal scales and highly accurate values for key material properties such as the mode-II interlaminar fracture toughness $G_{IIc}$.

Sun and co-workers [54] found that the measurement in the double cantilever beam (DCB) and the end-notch flexure (ENF) test was inaccurate because there was a significant absence of the “K-dominance zone” at interlaminar crack tips of these thin interlaminar fracture specimens. These tests often employ simple beam theory to characterize the complicated interlaminar fracture toughnesses. On the other hand, ENF beam bending tests are unable to eliminate the friction between cracked faces behind the crack tip. This phenomenon causes significant over-estimation of the $G_{IIc}$ values. Also, $G_{IIc}$ shows a wide variation of the test types (other than ENF). Such inaccuracies in measured material properties prevented high-quality simulations—especially the simulations of impact-induced delamination, which is mainly a mode-II-dominated fracture [60]. Because of the general lack of close agreement between impact experiments and simulations, simplified approaches were deemed helpful [44]. Among the variety of typical damage modes, matrix cracking usually produces only very low energy dissipation, and thus damage modeling/simulation should focus on other damage modes such as interlaminar damage (delamination) and fiber fracture [2]. Moreover, fiber fracture is rarely observed in low-energy impact tests. Therefore, this paper presents a simplified upper-bound approach for predicting the delamination areas. This approach will rely on the through-thickness Young’s modulus of a composite system, which is nominally difficult to measure [16]. Hence, we introduce a nanoindentation approach to measure this particular Young’s modulus.

Nanoindentation is an efficient technique for measuring certain mechanical properties of materials and devices at small length scales [41,37,42,25,10,12,18,55,17,58]. The mechanical foundation of nanoindentation is the indentation law connecting the indentation force and deformation. Previous research showed that at the beginning of an indentation process (elastic indentation), classical Hertz’s contact law is applicable to both short time-scale impact (microseconds) and small length-scale nanoindentation (nanometers) [41,2]. Therefore, we can build a correlation between nanoindentation and out-of-plane impact using multiscale indentation mechanics theory by conducting impact and nanoindentation experiments for the same material system. Indeed, a number of researchers did pioneering work and employed static macro/micro-indentation to understand the analogous projectile impact problem [2]. The major difference is that the current nanoindentation approach yields more accurate measurements, which are particularly important to predicting impact damage.

Based on the above analysis, our proposed approach is schematically displayed in Fig. 1 and consists of three steps. (A) Nanoindentation tests are used to measure the mechanical properties of the target materials. (B) The nanoindentation data are utilized in computational-mechanics-based predictions. (C) Impact experiments are conducted to validate the prediction results. The damage predicted from our method serves as an upper-bound that can be used for preliminary designs of composites, while more accurate simulations may be conducted later.

2. Methodology

In this section, first, the principles of indentation mechanics of isotropic and orthotropic materials are established to extract a key material property. Next, a connection between nanoindentation and impact mechanics is established for obtaining the maximum impact force. This is followed by the description of a computational method for modeling composite damage. Finally, measurements from nanoindentation and impact experiments are presented, and their results are compared with the results from the proposed prediction method for validation.

2.1. Indentation mechanics for isotropic materials and orthotropic composite materials

As shown in Fig. 2, the load $P$ of a spherical indenter is a function of the elastic indentation depth $h$ and the indenter radius $R$ based on Hertz’s contact law [20]:

$$P = \frac{4}{3}\sqrt{E_r h^3} = C_{NI} h^2$$

where $C_{NI}$ is the contact stiffness of nanoindentation, the reduced modulus $E_r$ is determined by the Young’s modulus $E$ and the Poisson’s ratio $\nu$ between the isotropic and homogenous target and indenter materials, and the subscript $i$ refers to the indenter:

$$\frac{1}{E_r} = \frac{1}{E_i} + \frac{1 - \nu_i^2}{E_i}$$

For orthotropic fibrous composite laminates (Fig. 2), a 1-2-3 material coordinate system is often employed. Here, 1 (or x) refers to the major fiber direction, 2 (or y) refers to the transverse in-plane direction or the minor fiber direction (fewer fibers compared to the 1-direction), and 3 (or z) refers to the thickness direction. Because the indenter is perpendicular to the composite laminate, the measured Young’s modulus is the through-thickness Young’s modulus $E_T$. Based on the classical work by Willis [56], Hertz’s contact law is applicable to the elastic static or dynamic indentation of orthotropic composite materials.

Recently, Xu et al. [64,63] employed an approximate expression of the reduced modulus as a function of $E_T$ because no sophisticated models are used for current composite indentation:

![Fig. 1. Framework for combining nanoindentation and computational mechanics simulations for assessing damage in composite material systems.](image-url)
2.2. Maximum impact force as a function of contact stiffness and impact energy

Abrate [2] and Andrews et al. [4] applied contact mechanics principles to moving spherical and sharp indenters/projectiles. They found that the maximum impact force $P_{\text{max}}$ was determined by the impact energy of the projectile ($W$)

$$ P_{\text{max}} = \frac{\lambda}{3} W C_{\text{IP}}^2, \quad (4) $$

where $\lambda$ is a constant related to the projectile shape but independent of the boundary conditions. $C_{\text{IP}}$ is the contact stiffness of impact between the projectile and the target, and it is calculated based on Hertz’s law:

$$ C_{\text{IP}} = \frac{4}{3} \sqrt{R_{\text{IP}} E_i}, \quad (5) $$

where $R_{\text{IP}}$ refers to the radius of the projectile with a spherical head, which is the only shape considered in this study. Hertz’s law is accurate only for elastic spherical indentation cases.

2.3. Multiscale indentation mechanics for nanoindentation and impact of the same material

Based on Eqs. (1) and (5), $C_{\text{IP}}$ can be obtained from the contact stiffness of nanoindentation $C_{\text{NI}}$ in the same target material system,

$$ C_{\text{IP}} = C_{\text{NI}} \sqrt{R_{\text{IP}} / R_{\text{NI}}} \left[ E_{\text{IP}}^3 / E_{\text{NI}}^3 \right], \quad (6) $$

where $E_{\text{NI}}$ and $E_{\text{IP}}$ are the reduced moduli of nanoindentation and impact, respectively. A few materials show the strain rate effect on their Young’s moduli below 5 GPa. Meanwhile, in most nanoindentation experiments, the diamond nanoindenter has a Young’s modulus of 1,000 GPa or higher, while that of a steel projectile used in the impact test is 200 GPa. Thus, a simplified relation between these two reduced moduli for the same target material was obtained based on Eq. (6). For the current carbon fiber/epoxy composite laminates, the relation is $C_{\text{IP}} \approx 0.97 C_{\text{NI}} \sqrt{R_{\text{IP}} / R_{\text{NI}}}$ after the approximate value of $E_{\text{IP}}^3 / E_{\text{NI}}^3$ is determined. The coefficient of 0.97 was calculated for the current composite system (taking into account $E_3$ of the current composite system (please, see Table 1); E and $v$ of steel impactor ($E = 200$ GPa, $v = 0.3$) and diamond nanoindenter ($E = 1000$ GPa, $v = 0.2$) [63]), while different values are expected for other material systems. Therefore, a multiscale relation is formed between the two radii of the nanoindenter $R_{\text{NI}}$ and the projectile $R_{\text{IP}}$ (often on the scales of micrometer vs. millimeter). Therefore, a multiscale indentation mechanics approach is utilized to connect the two very different events of impact and nanoindentation as follows:

1. The contact stiffness of nanoindentation $C_{\text{NI}}$ is obtained by fitting Eq. (1), and the contact stiffness of impact $C_{\text{IP}}$ is calculated.
2. The maximum impact force is predicted using $C_{\text{IP}}$ and the impact energy in Eq. (4).
3. The impact damage is predicted using the maximum equivalent impact force obtained from step 2), and then the results are validated by impact experiments.

Hence, we develop a general indentation mechanics approach by simply measuring the Young’s modulus $E_3$ as the initial task. It should also be noted here that the equivalent impact force includes not only the key Young’s modulus value but also the contact mechanics information such as the projectile’s radius and impact energy. Hence, it provides sufficient information for the predictions.

The key highlight of the proposed approach is its inclusion of the local dynamic indentation deformation while ignoring the global bending/shear deformation during a low-energy impact event. Therefore, this approach is employed to predict the upper-bound of damage approximately 0.03–10 N/s. It should be noticed that other authors employed the cylinder specimens under a 2-D strain state to measure the strain rate effect of stiffness using Hopkinson bars, while our indentation test yields a complicated 3-D strain state of the flat laminates. Quinn’s work was very close to our work so we mainly accepted their conclusion. Therefore, unless the strain rate effect on the Young’s moduli was reported for the material system, Eq. (6) is not related to the strain rate as a simplified result. On the other hand, the strain rate effect will be considered for the dynamic strengths and fracture toughnesses during the impact damage prediction which is described in Section 4, if these data are available.

Common polymeric matrices such as epoxy tend to have their Young’s moduli below 5 GPa. Since the through-thickness modulus of the corresponding composite systems are mainly controlled by their matrices [16], their effective values are usually less than 20 GPa. Meanwhile, in most nanoindentation experiments, the diamond nanoindenter has a Young’s modulus of 1,000 GPa or higher, while that of a steel projectile used in the impact test is 200 GPa. Thus, a simplified relation between these two reduced moduli for the same target material was obtained based on Eq. (6). For the current carbon fiber/epoxy composite laminates, the relation is $C_{\text{IP}} \approx 0.97 C_{\text{NI}} \sqrt{R_{\text{IP}} / R_{\text{NI}}}$ after the approximate value of $E_{\text{IP}}^3 / E_{\text{NI}}^3$ is determined. The coefficient of 0.97 was calculated for the current composite system (taking into account $E_3$ of the current composite system (please, see Table 1); E and $v$ of steel impactor ($E = 200$ GPa, $v = 0.3$) and diamond nanoindenter ($E = 1000$ GPa, $v = 0.2$) [63]), while different values are expected for other material systems. Therefore, a multiscale relation is formed between the two radii of the nanoindenter $R_{\text{NI}}$ and the projectile $R_{\text{IP}}$ (often on the scales of micrometer vs. millimeter). Therefore, a multiscale indentation mechanics approach is utilized to connect the two very different events of impact and nanoindentation as follows:

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The key highlight of the proposed approach is its inclusion of the local dynamic indentation deformation while ignoring the global bending/shear deformation during a low-energy impact event. Therefore, this approach is employed to predict the upper-bound of damage.
rather than the actual damage itself. It is similar to the common mechanics of materials approach of calculating the maximum impact stress of a slender beam subjected to projectile impact (one-point impact). When the projectile stops, the kinetic energy of the projectile is fully converted to the strain energy of the beam, and a statically equivalent impact force is obtained [6]. Then, the maximum impact stress can be easily calculated using classical beam theory. Obviously, the local dynamic indentation, which is within the theory of elasticity, is not considered in the mechanics of materials approach. Hence, one advantage of our proposed approach is simplifying the prediction of impact damage, and the results will be validated by our new and past impact tests.

Before this study, we conducted similar nanoindentation and drop-weight impact experiments on hard polymers and two different polymeric matrix composites (glass/carbon fiber reinforced vinyl ester) to validate the maximum impact force prediction using Eqs. (4)-(6) under the impact energy from 1 to 5 J [65]. As expected, the predicted maximum impact force was always higher than the measured one because the proposed approach only included the indentation deformation. The major difference between our current and previous studies is that no damage prediction was conducted in the previous cases. Moreover, these composite material systems, such as the polymeric matrix systems, were different between our current and previous studies.

2.4. Modelling the composite damage

The present composite damage model considers both intra- and interlaminar damage, which is based on other researchers’ work. The intra-laminar damage is modeled through a 3-D composite damage model based on the generalization of Hashin’s quadratic failure criteria [26] and Puck’s action plane model [45]. The model is coded as a VUMAT user subroutine for Abaqus/Explicit based on the lecture notes [1] and the investigations of Pederson [43] and Nie [39]. In this model, the tensile/compressive matrix and fiber failure modes are treated separately, and the calculation starts with computing the initial or undamaged orthotropic elastic parameters \( C_{ij} \) (i, j = 1, 2, 3). Then, the elastic parameters \( C_{ij} \) in the presence of damage are expressed as

\[
\begin{align*}
C_{11} &= (1 - d_f)C_{11}^0 - d_fE_{11}(1 - \nu_{23}\nu_{13})T, \\
C_{22} &= (1 - d_f)(1 - d_m)C_{22}^0 - (1 - d_f)d_mE_{22}(1 - \nu_{12}\nu_{31})T, \\
C_{33} &= (1 - d_f)(1 - d_m)C_{33}^0 - (1 - d_f)d_mE_{33}(1 - \nu_{13}\nu_{21})T, \\
C_{12} &= (1 - d_f)(1 - d_m)C_{12}^0 - (1 - d_f)d_mE_{12}(1 - \nu_{13}\nu_{21})T, \\
C_{23} &= (1 - d_f)(1 - d_m)C_{23}^0 - (1 - d_f)d_mE_{23}(1 - \nu_{12}\nu_{31})T, \\
C_{13} &= (1 - d_f)(1 - d_m)C_{13}^0 - (1 - d_f)d_mE_{13}(1 - \nu_{23}\nu_{13})T, \\
G_{12} &= (1 - d_f)(1 - d_m)G_{12}^0, \\
G_{13} &= (1 - d_f)(1 - d_m)G_{13}^0, \\
G_{23} &= (1 - d_f)(1 - d_m)G_{23}^0, \\
\Gamma &= 1/(1 - \nu_{12}\nu_{21} - \nu_{13}\nu_{31} - 2\nu_{23}\nu_{13} - 2\nu_{12}\nu_{31}),
\end{align*}
\]

for which the global fiber failure \( d_f \) and the matrix failure \( d_m \) are

\[
\begin{align*}
d_f &= 1 - (1 - d_m)(1 - d_f), \\
d_m &= 1 - (1 - d_f)(1 - d_m).
\end{align*}
\]

In Eqs. (7) and (8), \( d_f, d_m, d_m, \) and \( d_m \) are the local damage variables for the tensile/compressive fiber and tensile/compressive matrix failure modes, respectively. Therefore, combining Hashin’s quadratic failure criteria for the fiber damage and Puck’s action plane model for the matrix damage, these local damage variables are expressed as

Tensile fiber mode: \( \left( \frac{\sigma_{11}}{X_{11}} \right)^2 + \left( \frac{\sigma_{12}}{X_{12}} \right)^2 + \left( \frac{\sigma_{13}}{X_{13}} \right)^2 = 1, \quad d_f = 1, \) \hspace{1cm} (9)

Compressive fiber mode: \( \left( \frac{\sigma_{11}}{X_{11}} \right)^2 = 1, \quad d_f = 1, \) \hspace{1cm} (10)

Tensile and compressive matrix mode:

\[
\begin{align*}
\sigma_{11}^2 + \left( \frac{\sigma_{12}}{X_{12}} \right)^2 + \left( \frac{\sigma_{13}}{X_{13}} \right)^2 &= 1, & \quad d_m = 1, \\
\sigma_{12}^2 + \left( \frac{\sigma_{13}}{X_{13}} \right)^2 &= 1, \quad d_m = 1, \\
\sigma_{13}^2 + \left( \frac{\sigma_{12}}{X_{12}} \right)^2 &= 1, \quad d_m = 1.
\end{align*}
\]

Table 1

<table>
<thead>
<tr>
<th>Parameters obtained from the literature</th>
<th>Values</th>
</tr>
</thead>
<tbody>
<tr>
<td>Tensile longitudinal modulus ( E_1 ) (GPa)</td>
<td>[14] 164</td>
</tr>
<tr>
<td>Tensile transverse modulus ( E_2 ) and ( E_3 ) (GPa)</td>
<td>[27] 8.98</td>
</tr>
<tr>
<td>Shear modulus ( G_{12} ) (GPa)</td>
<td>[14] 5.02</td>
</tr>
<tr>
<td>Shear moduli ( G_{13} ) and ( G_{23} ) (GPa)</td>
<td>[14] 3.0</td>
</tr>
<tr>
<td>Poisson’s ratio ( \nu_{12} ), ( \nu_{13} )</td>
<td>[14] 0.32</td>
</tr>
<tr>
<td>Poisson’s ratio ( \nu_{23} )</td>
<td>[14] 0.49</td>
</tr>
<tr>
<td>Tensile longitudinal strength ( X_{11} ) (MPa)</td>
<td>[27] 2905</td>
</tr>
<tr>
<td>Compressive longitudinal strength ( X_{13} ) (MPa)</td>
<td>[27] 1569</td>
</tr>
<tr>
<td>Tensile transverse strength ( X_{22} ) (MPa)</td>
<td>[27] 78.9</td>
</tr>
<tr>
<td>Compressive transverse strength ( X_{23} ) (MPa)</td>
<td>[27] 248</td>
</tr>
<tr>
<td>Calibrated shear strength ( X_{23} ), ( X_{31} ) (MPa)</td>
<td>[27] 80</td>
</tr>
<tr>
<td>Interlaminar normal penalty stiffness ( K_n ) (GPa/mm)</td>
<td>[71] 1.2</td>
</tr>
<tr>
<td>Interlaminar shear penalty stiffnesses ( K_{12}, K_{13}, K_{23} ) (GPa/mm)</td>
<td>[71] 1.2</td>
</tr>
<tr>
<td>Mode I interlaminar fracture toughness ( G_{IC} ) (N/mm)</td>
<td>[14] 1.056</td>
</tr>
<tr>
<td>Mode II interlaminar fracture toughness ( G_{IIA}, G_{IIBC} ) (N/mm)</td>
<td>[14] 0.64</td>
</tr>
<tr>
<td>Interlaminar normal strength ( N ) (MPa)</td>
<td>[71] 68</td>
</tr>
<tr>
<td>Interlaminar shear strengths ( S ), ( T ) (MPa)</td>
<td>[71] 68</td>
</tr>
</tbody>
</table>

The interlaminar damage modeling is based on the cohesive behavior between the plies of the composite laminate available for the Abaqus/Explicit solver. Damage initiation is governed by the quadratic-traction-separation law

\[
\left( \frac{\Gamma}{N} \right)^2 + \left( \frac{\Gamma}{X} \right)^2 + \left( \frac{\Gamma}{T} \right)^2 \geq 1,
\]

for which \( \epsilon_n \), \( \epsilon_t \), and \( \epsilon_s \) are the interlaminar stresses dependent on the penalty stiffnesses \( K_n \), \( K_f \), and \( K_s \) in the normal and two shear directions; and \( N, S, \) and \( T \) are the interlaminar normal and shear strengths, respectively. In this expression, no irreversible deformation occurs till the tractions (or the interlaminar stresses) reach their peak values—i.e., initiation of damage. Thereafter, the cohesion between the plies will decrease with decreasing traction and increasing separation. During this phase, damage evolution occurs, which describes the degradation of cohesive stiffness between the plies. Here, a mixed-mode fracture criterion is used to describe the dependency of interlaminar fracture modes (mode I: opening, mode II: in-plane shear, mode III: out-of-plane shear) so that

\[
\frac{G_{IC}}{G_{IC}^0} + \frac{G_{IIA}}{G_{IIA}^0} + \frac{G_{IIBC}}{G_{IIBC}^0} \geq 1,
\]

where \( G_{IC}^0 \), \( G_{IIA}^0 \), and \( G_{IIBC}^0 \) are the interlaminar fracture toughness for the mode I, II, and III, respectively.
where $G_{II}$ and $G_{III}$ are the interlaminar energy release rates under modes I, II, and III, respectively, while $G_{IC}$, $G_{IBC}$, and $G_{IIBC}$ are the interlaminar fracture toughnesses. Since the present prediction approach focuses on low-energy impact, the strain rate effect is neglected. Moreover, no material data including the strain rate were available for the current composite system.

3. Experimental studies

3.1. Nanoindentation experiments for four kinds of aerospace composite laminates

The composite material system tested was IM7/977-3 carbon fiber/epoxy provided by the US Air Force Research Laboratory. Each of these specimens had 24 plies in one of four different stacking sequences: panel A $[-45/90/45/90/-45/0/45/90/45]_s$, panel B $[-45/45/-45/0/45/90/-45/45/-45/0]_s$, panel C $[-45/90/45/90/0]_3s$, and panel D $[-45/90/45/0]_3s$. Their average fiber volume percent was $68.5\%$ [22]. The size of each nanoindentation specimen was $10 \, \text{mm} \times 10 \, \text{mm} \times 3.2 \, \text{mm}$ (thickness). All eight specimens (two specimens cut from each panel) were grinded and polished to create a fine surface. All nanoindentation tests were conducted using an I-Micro nanoindentation system (Nano Mechanics, INC) with a spherical diamond indenter (radius $R_{nm} = 200 \, \text{µm}$). In order to reduce the inhomogeneous feature of the composite measurement, a spherical indenter with a large radius (at least $100 \, \text{µm}$) is recommended. During the nanoindentation tests, each specimen incurred $10 \times 10$ indents near its center, and two adjacent indents were separated by $70 \, \text{µm}$ in the x and y directions. The maximum applied load was low to ensure no obvious plastic deformation during the indentation process. The maximum indentation depth was kept below $350 \, \text{nm}$ to ensure elastic deformation. Since the average diameter of a carbon fiber is around $10 \, \text{µm}$ [16], the edge length of a square composite representative volume element (RVE, based on $68.5\%$ fiber volume) was around $10.7 \, \text{µm}$. Given the large size of the spherical indenter, our results were less sensitive to random material inhomogeneity. Moreover, we made 100 indentations on each composite specimen. The maximum indentation depth is an important parameter. Deep indentations may cause local permanent deformation or damage, while in very shallow indentations, the RVE might not fully deform, so the elastic constant cannot be measured. We suggest that $2\%$–$3\%$ of the edge length of a square RVE is the maximum indentation depth so that the RVE fully deforms. Therefore, our indentation approach not only yields purely elastic deformation but also reduces the effect of material inhomogeneities. The maximum indentation depth without significant plastic deformation was validated by comparing the load/indentation depth curves of panel A. As shown in Fig. 3, the unloading path was quite close to the loading path, and permanent deformation was not clearly seen. Hence, we fitted the contact stiffness of nanoindentation $C_{np}$ of each panel using Eq. (1), and the through-thickness Young’s moduli of four composite panels [66] are $8.33 \pm 0.26 \, \text{GPa}$ (A), $8.72 \pm 0.31 \, \text{GPa}$ (B), $9.40 \pm 0.38 \, \text{GPa}$ (C), $10.18 \pm 0.37 \, \text{GPa}$ (D).

If the fiber volume fraction is low or a sharp indenter is employed, the measured reduced moduli of the fiber or matrix are very different. So, we cannot measure the Young’s modulus of a composite specimen, which should be one value to represent the global deformation of the matrix and the fiber. Recently, based on micromechanics, we can measure the lower bounds of the Young’s moduli for the above cases [64].

3.2. Out-of-plane impact experiments of the same composite laminates

The dimensions of all impact specimens were $101.6 \, \text{mm}$ (width) $\times$ $152.4 \, \text{mm}$ (length) $\times$ $3.2 \, \text{mm}$ (thickness) according to ASTM D7136 standard for composite impact experiments. All specimens (six for each panel except panel B) were tested under the same impact energy of $5 \, \text{J}$. Before the experiments, all specimens were C-scanned to determine if there were any initial defects. The composite specimens were clamped onto a fixture with a round hole of $76.2 \, \text{mm}$ diameter. This impact fixture was fixed to a steel base inside a DYNATUP drop-weight tower. The impact experiment was conducted using a $15.9 \, \text{mm}$ (5/8 in.) diameter hemispherical upper (mass: $3.37 \, \text{kg}$), which hit the center of the specimen, and the internal damage was examined by C-scanning again. More details of these impact experiments were reported previously by one of the authors [22].

4. Numerical simulation procedures

Based on the stacking sequences of four panels, the corresponding material orientations were assigned to each ply, and four edges of the composite laminate were clamped as seen in Fig. 4. All mechanical properties of the IM7/977-3 composite material are listed in Table 1. Most of these data were measured at the Air Force Research Laboratory in Ohio. The solution domain for the composite laminates was partitioned. A higher mesh density was implemented in the vicinity of the impact site in order to eliminate the convergence problems that may arise from surface-based cohesion, while a lower mesh density was used in the rest of the solution domain for the sake of computational efficiency. Each ply was discretized along its thickness with single C3D8R linear hexahedral reduced integration elements. The main selection criteria for the C3D8R reduced integration elements—i.e., one integration point per element—were to decrease the computational cost and prevent shear or volumetric locking, which was the case for first-order fully integrated elements subjected to bending [21]. However, it is likely that severe mesh distortions may exist for the elements even though strain at the integration point is arbitrarily small or zero, which is known as the hourglass effect. To prevent such deformations, default hourglass control provided in Abaqus/Explicit was used by a relax stiffness method using the integral viscoelastic form defining anti-hourglass forces.

For each generated solution domain, the characteristic length of the element in the impact region was determined to be $0.625 \, \text{mm}$ following the mesh sensitivity analysis of Pederson [43]. For the outer regions, an element size of $2 \, \text{mm}$ was selected. Each simulation used full-scale models depicted in Fig. 4 due to the non-symmetric damage evolution observed in previous experimental studies. When the intra-laminar damage criteria of Eq. (8) were reached, stresses at the integration points were computed with reduced stiffness matrices.

In consideration to the interlaminar damage, general contact interactions with cohesive surfaces, which use tracking algorithms to ensure proper contact conditions to all existing interfaces, were applied. The quadratic-traction-separation law of Eq. (13) was used to predict damage initiation. Thereafter, post-damage behavior was treated as progressive stiffness degradation representing delamination with the
increasing damage, and it was numerically handled with the mixed-mode energy release rate criterion \cite{39}. As illustrated in Fig. 4(a), the tupper (a specific projectile used in drop-weight tests) was modeled with R3D4 rigid quadrilateral elements and constrained to move only towards the panel surface along the Z-axis without any rotational inertia. This was realized through assigning a rigid body reference point, through which the prescribed velocities of dynamic simulations and external loads of equivalent force simulations were applied to the tupper. A distance of 0.2 mm was initially set between the tupper and panel surfaces. The interaction between the tupper and the panel was defined for all existing interfaces with the Abaqus built-in general contact interaction algorithm and hard contact model with a user-defined viscous damping coefficient of 0.01 \cite{51}. This approach is able to minimize the oscillations arising from rigid body motions in the contact events, which were also reported in the literature \cite{49;44}. A Coulomb friction coefficient of 0.1 was applied between the tupper and the panel during the simulations based on sensitivity analysis in the literature \cite{52}.

Two types of numerical simulations, which were (1) dynamic and (2) equivalent force simulations (i.e., quasi-static), were conducted through central-difference time integration using Abaqus/Explicit and a VUMAT user subroutine \cite{29}. In the latter, a stable time increment was chosen so that the ratio of the kinetic energy of the laminates to the internal energy throughout the simulation was kept small (typically 5–10%) as recommended in Abaqus user documentation. This ensured that the strain rates and inertial forces were negligible. The dynamic simulation was mainly employed to check the prediction accuracy of the present damage models by comparing the simulation with the experimental load history. The equivalent force approach was used to predict the interlaminar damage only. For each time step, the strain increment was used as the input of the constitutive material model coded in the subroutine, and the stress and the corresponding damage states were obtained. Due to the large number of increments, all the simulations were run with double precision.

The key assumption of our approach is the simplified energy balance principle \cite{2}, because the kinetic energy of the projectile is converted into other energy forms:

\[
E_k = E_{\text{in}} + E_{\text{gb}} + E_d + E_f
\]  

(15)

where \(E_{\text{in}}\) is the strain energy caused by local projectile indentation, \(E_{\text{gb}}\) is the strain energy caused by global plate deformation such as bending, \(E_d\) is the energy dissipated in damage creation, and \(E_f\) is the friction energy (which is very small for a smooth projectile surface). In the equivalent force approach, \(E_{\text{in}}\) is simulated and included but \(E_{\text{gb}}\) is...
neglected, so extra energy from $E_{gb}$ virtually goes to $E_d$ when the total energy is fixed. As a result, the predicted damage should be larger than the actual damage, i.e., an upper-bound prediction. On the other hand, if the strain energy caused by global plate deformation is comparable with the strain energy caused by local indentation, a large prediction error is expected, and the proposed prediction will not work well.

5. Results and discussion

First, the dynamic simulation was validated with the impact tests conducted at 5 J energy in this study. Only the measured and predicted impact force histories were compared. Predicting damage was not conducted because of the concern of inaccurate material properties. Thereafter, the proposed equivalent force approach was employed to predict the delamination sizes in the current and previous (energy: 10 J) impact tests of the same specimens.

5.1. Validation of the dynamic simulation by impact load history

Dynamic simulations were conducted for the current impact tests for a total time period of $t = 5$ ms with a prescribed impact velocity of 1.72 m/s. For each time step, the contact forces between the tupper and the panels were computed, and the load history of each panel was generated and compared with the experimental data. In Fig. 5, all experiments were highly repeatable with very similar load curves. The average peak loads for panels A, B, C, and D were 3.76, 3.95, 3.27, and 3.87 kN, respectively, as listed in Table 2. The time to reach the maximum impact force was around 1.5 ms for all three panels except panel C. The simulated load history had similar trends with the experimental ones, although the simulated peak loads (4.51, 4.88, 3.98, and 4.47 kN, respectively) were higher than the measured values by 13–19%. In contrast to the peak loads, the simulated maximum tupper displacements were less than the experimental ones, with a percentage difference of 8%, 0.5%, 20%, and 9%, respectively. These discrepancies are attributed to some inaccurate material constants, e.g., the interlaminar penalty stiffness/strength values that were not directly measured but rather modified from literature [71].

It is also important to note that, after the onset of tupper-panel contact, the kinetic energy of the tupper was gradually transferred to the panels as strain energy, damage initiation and propagation, and fluctuations as captured in the load history plots. Especially, before the load reached its peak value, there were fluctuations that were presumably caused by the combined effects of intra- and interlaminar damage initiation, and evolution and strong vibrations in the initial tupper-panel

![Fig. 5. Panels subjected to an impact energy of 5 J: (a) Simulation setup and half-cut showing the simulated interlaminar damage evolution of Panel A (legend bar shows the damage: blue color refers to completely failed/damaged surfaces, gray color designates undamaged entities, and other colors denote partially damaged entities), (b) Comparison of impact load history for panels A–D between experiments and numerical simulations.](image-url)
contact [22].

The same issue was also observed previously by other researchers, and various oscillation damping-out solutions were proposed [49;44]. After the load reached its peak value, strain energy of the panel caused the tupper to bounce back, which is shown at $t = 3$ and $4$ ms in Fig. 5. In the recovery period, the impact load gradually decreased, while impact damage inside the panels did not change. The simulation shows that impact damage occurred even at the low impact energy of 5 J.

5.2. Upper-bound predictions of the maximum delamination sizes

Prediction results using the equivalent force approach were compared with the previous and current experimental results for the same composite panels of A–D. For this purpose, Eq. (4) was used to determine $P_{\text{max}}$ in simulations with a total duration $t = 2000$ ms. The $\lambda$ value was taken to be 0.38 (equivalent to $P_{\text{max}} \sim 5700$ N obtained from the previous 10 J impact test). The same $\lambda$ value was used for the current 5 J impact test using the identical composite laminates and the same ASTM testing standard. According to Abrate [2], $\lambda = 1.73$ for a spherical projectile on an isotropic material target. Now the target is a composite laminate in this study, we have to assume one $\lambda$ value. Moreover, previous experimental studies showed that the measured impact force was always lower than the predicted force using $\lambda = 1.73$ because this value was based on the 100% kinetic energy transfer assumption to the target. However, in our drop-weight impact experiments, significant projectile rebound (only partial kinetic energy transfer) was observed so $\lambda$ must be much smaller than 1.0. Here, we assume a reasonable $\lambda$ value based on the previous impact tests with impact energy of 10 J, and predict the current impact damage tests of impact energy of 5 J. So, our approach is a reasonable effort to explore a complicated mechanics problem.

The maximum interlaminar damage areas from the upper-bound predictions and experiments are compared in Table 3 and presented in Figs. 6 and 7. For each stacking sequence, 1–2 representative impact damage images are presented. From the predictions, only the damaged area representations, namely the nodal coordinates of the failed elements $\xi_j$, were extracted and fitted to the hull by means of Mathematica based on the reports by Weisstein [57] and Karakoç and Freund [30]:

$$C = \left\{ \sum_{j=1}^{N} \xi_j \cdot \text{crd} : \xi_j \geq 0 \hspace{1mm} \forall \hspace{1mm} j \text{ and } \sum_{j=1}^{N} \xi_j = 1 \right\}. \quad (16)$$

Thereafter, the damaged area was expressed as a planar non-self-intersecting polygonal area [7]. In Table 3, the predicted damage areas for panels A, B, and D are 13–26% larger than the ones experimentally measured in the 10 J impact tests. The same trend was observed at the impact energy of 5 J, but the difference between the predictions and measurements was smaller. This outcome was expected: because only indentation deformation was considered in the equivalent force approach while global bending/shear deformation was not included, and the predicted damage areas should be larger. Actually, a larger predicted damage area is better for the conservative design of damage tolerance, especially in the early design stage for composites. In our previous damage tolerance experiments, composite laminates with large impact damage (mainly delamination) had low residual compression strengths [62].

Panel C shows a large discrepancy between the predicted and measured damage areas, due to its low bending stiffness (or large bending deformation) of this particular stacking sequence [22]. Indeed, the stacking sequence of a composite laminate strongly affects its stiffness and failure [59,2]. Here, panel C differs significantly from the other panels, but this outcome does not contradict the proposed approach at all, instead, it exactly supports our assumption: the proposed approach is applicable to composite laminates/panels with small global plate deformation only. However, we are not able to obtain an applicable quantitative range for the proposed approach yet. If we simply employ the contact stiffness of impact CIP (which is related to indentation deformation) and the major bending stiffness of a composite laminate $D_{11}$ to compare indentation/bending deformation, the units of these two parameters are totally different or not comparable ($C_{IP} \sim N/m/1.5$, $D_{11} \sim Nm$). Fortunately, almost all current aerospace composite structures are much thicker than the four panels, and have small bending deformation according to one co-author at the US Air Force Research Laboratory. Therefore, the proposed approach could be employed as an efficient tool for preliminary damage tolerance evaluation.

When comparing two different approaches, the impact velocity, mass or energy, shape of the projectile should be inputted individually into the dynamic simulation. In contrast, the equivalent force approach directly applies the maximum impact force on the specimen, which simplifies the prediction. Another advantage of our approach is the reduced need for impact tests. First, nanoindenters are widely accessible. Second, nanoindentation experiments require only one researcher and can be run automatically, while impact tests often require more researchers and operation time (e.g., the alignment). Since we predict the upper-bound of the damage sizes, it is expected that the predicted sizes can be much larger than the measured damage size (not close).

The key part of the proposed approach, i.e., the maximum impact force prediction, was validated through the impact force values measured from two different impact experiments also. The average maximum impact force for panels A, B, and D measured in the two experiments was 3.86 kN at 5 J and 5.70 kN at 10 J, so the ratio was 1.44 between 10 J and 5 J impact energy levels. Because the two experiments used the same impact mechanics and material conditions (except the impact energy level), the predicted ratio of two maximum impact force values using Eq. (4) was 1.48, quite close to the measured value. Therefore, this is a unique pilot study for low-energy impact on composites validated by our current and past experiments. In the future, we plan to focus on composites subjected to high-energy impact. It is important to notice that we only employed standard impact/nanoindentation experiments, and composite damage modeling. The novelty of this paper is building the simplified connection between two extreme (fast/small) events.

### 6. Conclusions

The present pilot study introduced an equivalent impact force framework by using both indentation mechanics and computational mechanics to efficiently predict the upper-bound of interlaminar impact damage in composite laminates. Such framework is important since it simplifies the prediction of impact damage especially for the practical applications of thick composite laminates, which are widely used in aerospace structures. In order to demonstrate the framework,
nanoindentation (nanometer-scale—very small) and low-energy projectile impact (microsecond—very fast) experiments were carried out on the carbon fiber/epoxy composite laminates using a multiscale indentation mechanics relation. Even though some of the interfacial material properties had to be modified from the literature, the upper-bound prediction provided conservative and confident impact damage assessment with the minimum computational cost.

CRediT authorship contribution statement

L. Roy Xu: Conceptualization, Methodology, Investigation, Project administration, Funding acquisition. Alp Karakoğlu: Methodology, Investigation, Validation. Mark Flores: Investigation, Validation. Haibin Ning: Investigation. Ertugrul Taciroglu: Conceptualization, Methodology, Supervision, Funding acquisition.

Declaration of Competing Interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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Data Availability Statement

The raw/processed data required to reproduce these findings cannot...
be shared at this time due to technical or time limitations by the US Air Force.

Appendix A. Supplementary material

Supplementary data to this article can be found online at https://doi.org/10.1016/j.compstruct.2021.115110.

References

